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Gun Tube Coatings in Distress

by Paul J. Conroy, Paul Weinacht,
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ARL-TR-2393**February 2001**

Gun Tube Coatings in Distress

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Abstract

The effects of thermal, chemical, and mechanical stress components of erosion on the degradation and failure of current and potential future gun tube coating materials are investigated. Potential failure mechanisms at the interface are shown and discussed. The interfacial coating/base material stress calculation methodology, including coating crack formation and coating adhesion criteria, is presented. Possible erosion mitigation concepts are also presented to counteract the formation of cracks in the coating.

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1. Introduction

The thermochemical modeling description used for the calculations in this report has been previously reported (Bundy et al. 1996; Conroy et al. 1997, 1998, 1999a, 1999b). The current methodology has the addition of an ablative surface coating, which follows the same methodology as described in Weinacht and Conroy (1996). This report describes the potential stresses encountered by a contiguous coating where residual stresses from manufacturing have not caused cracking or delaminated from the substrate. Erosion calculations are presented for both chromium and tantalum coatings at the bottom of cracks, which form from thermal-mechanical cracking. A mechanistic description of the undermining and subsequent removal of the coating followed by discussions of the results is presented.

Historically, gun tubes were bare steel in the United States until World War II (WWII), when many coatings and types of steel were examined. These included Ta, Cr, Mo, W, Co, and stellite (Bush et al. 1946). As of August 1942 (Ritchie 1942), the only corporation in the United States to successfully plate a gun barrel with chromium was Battelle. Coating and erosion mitigation methods during the war, and shortly thereafter, involved primarily electroplating chromium or inserting a section of stellite into a .50-caliber machine gun. During the postwar years from 1946 to 1965, the literature on gun tube wear and erosion became scarce. Due to issues with weapon systems in Vietnam (namely the 175-mm gun as well as machine guns), a renewed interest in gun tube wear and erosion emerged around the mid-1960s. Meetings concerning liners, coatings, and additives took place (Ahmed and Picard 1970), and much research (Jones and Breitbart 1959; Frankle and Kruse 1967; Gehring and Meyer 1968; Grabke 1969; Hochrein and Thruvengadan 1973) was performed on the wear and erosion issue.

Mitigation efforts advanced from 1946 to the early 1980s with the advent of talc, waxes, and TiO_2 additives and with mixtures involving each (Brosseau and Ward 1975; Lannon and Ward 1981; Chen and Baldauf 1980; Ward and Brosseau 1980). The work from the mid-1960s through the early 1980s was summarized in two triservice meetings held at Picatinny Arsenal, NJ, by Jean Paul Picard in 1977

and 1982 (Picard and Ahmed 1977; Picard et al. 1982). From the mid-1980s through 1990, interest in gun tube wear and erosion waned due to fielded systems meeting mission requirements. Usually when an erosion issue would arise, as with the original M919 round which later became the "616W," a team was assembled to solve the problem with additives and propellant formulations (Talley and Owczarczak 1991). Currently, other than low-contraction (LC) electrodeposited chromium (Chen and Baldauf 1980) and the nitrided M242 25-mm barrel, there are no new fielded coatings or new methods of plating or coating a gun tube surface. However, there are over half a dozen ongoing projects in the U.S. Army ranging from functionally gradient ceramics to magnetron sputtering for coating/lining gun tubes. Despite these ongoing projects, it is still difficult to obtain a coating/liner which functions as well as the current methods.

Why is it so difficult to find or develop a good coating? Coatings mitigate erosion for two basic reasons. First, a successful coating is not reactive with the propellant product gases. Second, the coating serves to distribute and insulate the thermal load to the gun tube base material, thus reducing the effective peak temperature experienced by the gun steel. Typically, when the coating/lining material has both of these properties, the mechanical properties are not suitable for gun tubes; therefore, they are coatings or liners and usually not a significant structural load-bearing section of the tube. Another issue, somewhat separate from thermochemical erosion, is the wear due to the projectile passage, which implies that the coating must also be resistant to mechanical wear. Aside from these basic requirements, there are many other important details, such as coating adherence to the base material and the coefficient of thermal expansivity (CTE), which should be close to that of the base material to prevent thermal stress cracking and many other details. One final concern is that both the coating and the deposition technique must be cost effective (Montgomery 1999). When put together, the requirements for a coating are extremely demanding.

Describing a coating in an interior ballistic setting requires inclusion of the structural considerations. Therefore, the following structural derivation is provided to describe the coating crack distribution.

2. Structural Calculation Methodology

The coupled thermomechanical stress state of the barrel with a coating is required to determine the interfacial loads which contribute to coating breakdown. In order to solve for interfacial shear stresses between a coating and base material in an axisymmetric plane stress situation, the principal stresses must be computed for both coating and base material. It is important to note that the calculations contained within this report assume that the initial state of the tube and coating are unstressed. That is, residual stress due to plating and autofretting is not present. Inclusion of the nominal stress state is possible in future work. Following Timoshenko and Goodier (1970), note the following equation of equilibrium, where r , θ , and z denote radial, azimuthal, and axial directions.

$$\frac{\partial \sigma_r}{\partial r} + \frac{\sigma_r - \sigma_\theta}{r} = 0, \quad (1)$$

with the following Poisson relations for ϵ strains, with α as the CTE and ν the Poisson coefficient,

$$\epsilon_r - \alpha T = \frac{1}{E} (\sigma_r - \nu(\sigma_\theta + \sigma_z)), \quad (2)$$

$$\epsilon_\theta - \alpha T = \frac{1}{E} (\sigma_\theta - \nu(\sigma_r + \sigma_z)), \quad (3)$$

and

$$\epsilon_z - \alpha T = \frac{1}{E} (\sigma_z - \nu(\sigma_\theta + \sigma_r)). \quad (4)$$

For displacement u ,

$$\epsilon_r = \frac{\partial u}{\partial r}, \quad \epsilon_\theta = \frac{u}{r}, \quad \text{and} \quad \epsilon_z = \epsilon_o \quad (5)$$

provide a basis from which to compute the stress and displacement relations, assuming plane stress (i.e., $\sigma_z = 0$) reduces the stress components to

$$\sigma_\theta = \frac{E}{1-\nu^2} \epsilon_\theta + \nu \epsilon_r - (1+\nu)\alpha T, \quad (6)$$

$$\sigma_r = \frac{E}{1-\nu^2} \epsilon_r + \nu \epsilon_\theta - (1+\nu)\alpha T, \quad (7)$$

and

$$\sigma_z = 0. \quad (8)$$

The axial strain relationship reduces to

$$\epsilon_z = \alpha T - \frac{\nu}{E} \sigma_\theta + \sigma_r. \quad (9)$$

The following differential equations for the radial displacement result by substituting the previous stress functions (equations 6 and 7) into the equilibrium differential equation using the constitutive relations of equation 5:

$$\frac{\partial^2 u}{\partial r^2} + \frac{1}{r} \frac{\partial u}{\partial r} - \frac{u}{r^2} = \alpha(1+\nu) \frac{\partial \Delta T}{\partial r}, \quad (10)$$

or

$$\frac{\partial}{\partial r} \left(\frac{1}{r} \frac{\partial (ru)}{\partial r} \right) = \alpha(1+\nu) \frac{\partial \Delta T}{\partial r}, \quad (11)$$

which, when solved for u , results in

$$u = \alpha(1+\nu) \frac{1}{r} \int_a^r \Delta T dr + C_1 r + \frac{C_2}{r}. \quad (12)$$

Through the relations of equation 5, this, in turn, supplies the displacements ϵ_r and ϵ_θ as

$$\epsilon_r = -\frac{\alpha(1+\nu)}{r^2} \int_a^r \Delta T dr + \frac{\alpha(1+\nu)}{r} \Delta T + C_1 - \frac{C_2}{r^2} \quad (13)$$

and

$$\epsilon_\theta = \frac{\alpha(1+\nu)}{r^2} \int_a^r \Delta T dr + C_1 + \frac{C_2}{r^2}. \quad (14)$$

Using these results in stress relations (equations 7 and 8) produces the radial and hoop stress as

$$\sigma_r = \frac{E\alpha}{r^2} \int_{r_i}^r r \Delta T dr + \frac{E}{1-\nu^2} \left(C_1(1+\nu) - \frac{C_2}{r^2}(1-\nu) \right) \quad (15)$$

and

$$\sigma_\theta = \frac{E\alpha}{r^2} \int_{r_i}^r r \Delta T dr - E\alpha \Delta T + \frac{E}{1-\nu^2} \left(C_1(1+\nu) + \frac{C_2}{r^2}(1-\nu) \right), \quad (16)$$

where the constants C_1 and C_2 are found by applying the following boundary conditions at $r=r_i$ and $\sigma_r = -P_i$ and at $r=r_o$, $\sigma_r = -P_o$, resulting in

$$C_1 = \frac{C_2(1-\nu)}{r_i^2(1+\nu)} - \frac{P_i(1-\nu^2)}{E(1+\nu)} \quad (17)$$

and

$$C_2 = \frac{(1-\nu^2)}{E(1-\nu)} \left(P_i - P_o + \frac{E\alpha}{r_o^2} \int_{r_i}^{r_o} r \Delta T dr \right) \left(\frac{r_o^2 r_i^2}{r_o^2 - r_i^2} \right). \quad (18)$$

To compute the stresses for the coating (liner) and base material, the interfacial pressure must be known. The displacement compatibility at the interface supplies this interfacial pressure by stating that the displacement of the coating, subscripts c, must be equal to that of the base material, subscripts b, at the interface (i.e., from equation 12):

$$\alpha_c(1+\nu_c) \frac{1}{r_I} \int_{r_i}^{r_I} \Delta T dr + C_{1c} r_I + \frac{C_{2c}}{r_I} = \alpha_b(1+\nu_b) \frac{1}{r_I} \int_{r_i}^{r_o} \Delta T dr + C_{1b} r_I + \frac{C_{2b}}{r_I}, \quad (19)$$

where C_{1c} and C_{2c} are

$$C_{1c} = \frac{C_{2c}(1-\nu_c)}{r_i^2(1+\nu_c)} - \frac{P_i(1-\nu_c^2)}{E_c(1+\nu_c)} = A1c * C_{2c} - A2c * P_i, \quad (20)$$

and

$$C_{2c} = \frac{(1-\nu_c^2)}{E_c(1-\nu_c)} \left(P_i - P_I + \frac{E_c \alpha_c}{r_I^2} \int_{r_i}^{r_I} r \Delta T dr \right) \left(\frac{r_I^2 r_i^2}{r_I^2 - r_i^2} \right) = A1b * C_{2b} - A2b * P_I, \quad (21)$$

while C_{1b} and C_{2b} are

$$C_{1b} = \frac{C_{2b}(1-\nu_b)}{r_I^2(1+\nu_b)} - \frac{P_I(1-\nu_b^2)}{E_b(1+\nu_b)} = A1b * C_{2b} - A2b * P_I, \quad (22)$$

and

$$C_{2b} = \frac{(1-\nu_b^2)}{E_b(1-\nu_b)} \left(P_I - P_o + \frac{E_b \alpha_b}{r_I^2} \int_{r_I}^{r_o} r \Delta T dr \right) \left(\frac{r_I^2 r_o^2}{r_o^2 - r_I^2} \right) = A3b * P_I - A3b * P_o + A3b * A4b. \quad (23)$$

Substitution of the defined variables into the displacement compatibility relation of equation 19, along with the following variables for the integral terms,

$$AOc = \alpha_c(1+\nu_c) \frac{1}{r_I} \int_{r_I}^{r_o} r \Delta T dr, \quad (24)$$

and

$$AOB = \alpha_b(1+\nu_b) \frac{1}{r_I} \int_{r_I}^{r_o} r \Delta T dr = 0, \quad (25)$$

produces the interfacial stress, P_I , as

$$P_I = \frac{A2c * r_I * P_i - A3c * (P_i + A4c) \left(A1c * r_I + \frac{1}{r_I} \right) - A3b * (P_o + A4b) \left(A1b * r_I + \frac{1}{r_I} \right) - AOc}{A2b * r_I - A3c * \left(A1c * r_I + \frac{1}{r_I} \right) - A3b * \left(A1b * r_I + \frac{1}{r_I} \right)}. \quad (26)$$

Thus, substitution of equation 26 into C_1 and C_2 enables the calculation of the radial and hoop stress. The resultant difference between the hoop stress from the coating and base material results in the interfacial radial shear stress for contiguous materials. This analytical formulation was tested against ANSYS for constant inner pressure and temperature loads and against Lamé's analytical solutions for constant inner pressure loads, as presented in Figure 1.

An estimate of the axial interfacial shear stress follows from the computation of the axial strain for the coating and base material independently, with no axial stress, from equation 4 as

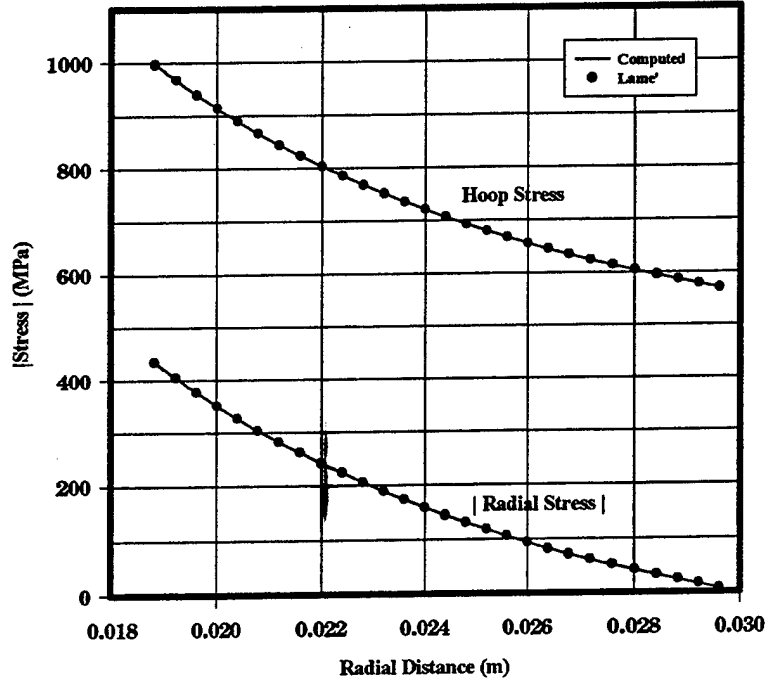


Figure 1. Thick-Walled Cylinder Radial and Shear Stress Plotted Radially With Constant 500-MPa Inner Wall Pressure, No Temperature Effects.

$$\epsilon_z = \alpha T - \frac{\nu}{E} \sigma_\theta + \sigma_r. \quad (27)$$

Substituting and reducing equations 15 and 16 into equation 27 for σ_θ and σ_r produces

$$\epsilon_z = 1 + \nu \left(\alpha \Delta T - \frac{2\nu C_1}{1 - \nu^2} \right), \quad (28)$$

where ΔT is the interfacial temperature increase. The equivalent axial stress associated with this axial strain is simply $\epsilon_z E$. The difference between the coating and base material's equivalent axial stress is the axial shear, which must be supported by the interface.

The possible distance between fractures of the coating can be represented by the balance of the shear stress, with the ultimate strength (UTS) of the coating axially as

$$\tau_{axial}(2\pi r_I dl) = UTS\pi(r_I^2 - r_i^2), \quad (29)$$

which reduces to the axial interfracture length of

$$dl_a = \frac{UTS\pi(r_I^2 - r_i^2)}{\tau_{axial}2\pi r_I}, \quad (30)$$

while radially, the interfracture length is simply

$$dl_r = \frac{UTS(r_I - r_i)}{\tau_r}. \quad (31)$$

Figure 2 shows interfacial shear stresses experienced during the firing of an armor-piercing, fin-stabilized, discarding-sabot (APFSDS) round in a 120-mm M256 cannon. Note that for high contractile (HC) chromium, the stresses never appear to exceed the ultimate values because HC chromium is very brittle. The shear stresses possibly exceed LC chromium's ultimate stress. However, note that LC chromium is more ductile than HC chromium and fails at about 10% strain. Tantalum shows much higher values of interfacial shear due to CTE differences. Tantalum also exhibits much lower values of ultimate strength. Tantalum, however, is extremely ductile (Montgomery 1999), which makes it highly unlikely to crack.

The coatings do not show the tendency to readily debond or fail in shear at the interface. If they did, then the coating would fail rather quickly and would not be successful, and the guns would not fire for hundreds of rounds. The coatings appear to relieve stresses through plastic strain or cracking. The minimum crack distances for a 120-mm M256 tank cannon firing an APFSDS round in which the barrel is coated with 0.01 in of either chromium or tantalum are presented in Table 1. Dilation effects were not taken into account in this determination. The larger distance between radial cracks suggests fewer radial cracks in the tube, causing more axial cracking. Dr. Paul Cote of Benet Weapons Laboratory (Watervliet Arsenal, Watervliet, NY) discussed some of the cracking that he observed between 155-mm and 120-mm systems as well as HC and LC chromium (Cote and

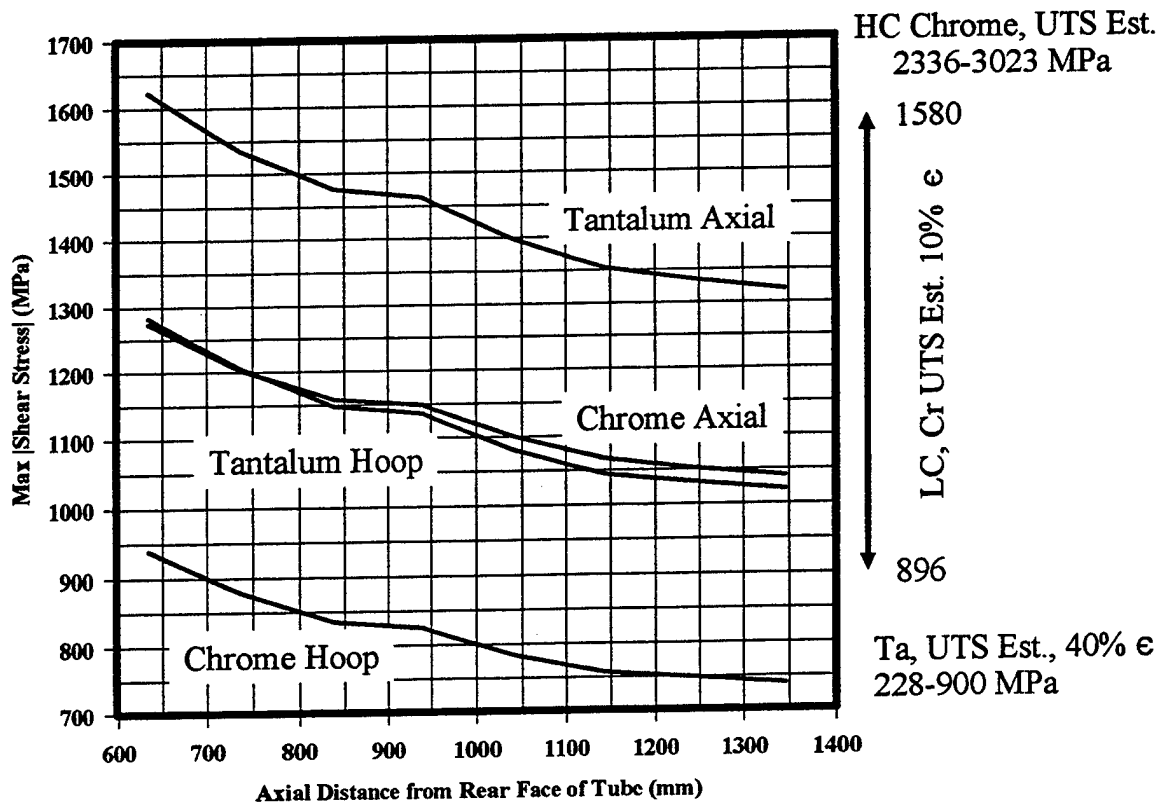


Figure 2. Maximum Computed Interfacial Shear Stresses for 0.010-in Chromium and Tantalum Coatings With UTS Values (Single APFSDS Round From an M256 120-mm Cannon).

Table 1. Computed Nominal Minimum Distance Between Cracks Not Considering Dilation Effects

| Coating | UTS Used in Calculation ^a (MPa) | Radial Distribution (m) | Axial Distribution (m) |
|----------|---|----------------------------|---------------------------|
| Chromium | 413 | 1.030e-4 | 7.532e-5 |
| Tantalum | 900 | 1.651e-4 | 1.291e-4 |

^aMatweb (1999).

Rickard 1999). According to Dr. Cote, there are microcracks caused by residual stresses created during the plating process. Macroscopic cracks arising from thermomechanical stresses also exist. The computed cracking is more of the macroscopic type of cracks found by Dr. Cote than the microcracks. These chromium-coating crack results are similar to those observed in gun tubes

(Sopok et al. 1988). Usually the dimensions of the coating “platelets” are similar to the coating thickness. These calculations were performed using nominal physical values which may not represent those of the actual as-deposited materials. Given all the intrinsic qualifications of these coating vs. coating calculations, tantalum should show less propensity than chromium for cracking. The fact that there are cracks in existing chromium coatings and mechanisms to produce these cracks leads to the following section in which the computation of erosion at the bottom of the cracks is performed.

3. Erosion Calculations at Coating Crack Base

Figure 3 represents what occurs beneath the coating near a crack. A jet of combustion gases washes through the crack, subsequently reacting and eroding the substrate material and leaving a pit with horizontal extensions near the interface. Many species of materials in the eroded region have been identified and include sulfur, oxygen, and talc, which are from the propellant additives (Cote and Rickard 1999; Fisher et al. 1982). Sulfur, which penetrates deeply into the eroded pits, comes from the black powder ignitors and potassium sulfide flash suppressants used in the charges.

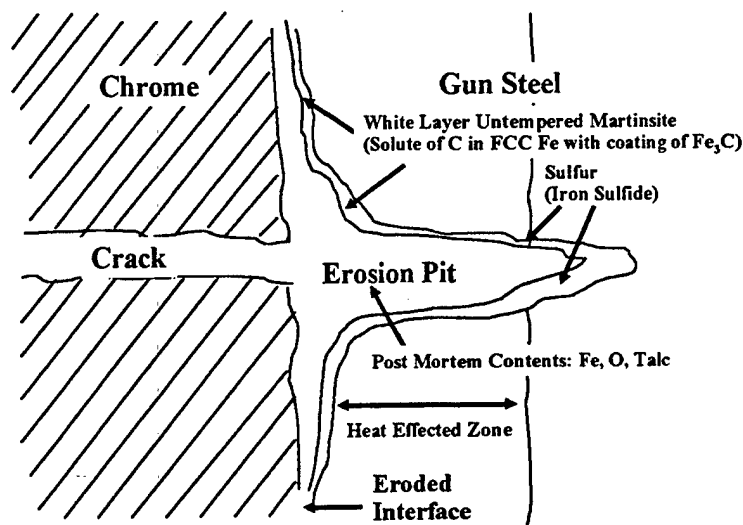


Figure 3. Description of Postmortem Erosion Pit (Source: Dr. Paul Cote, Benet Laboratories).

Figure 4 is a cartoon of the numerical simulation of the erosion at the bottom of a crack. It is important to realize that crack orientation is critical. If the crack is longitudinal, it would enable propellant product gases to wash directly in and out with tremendous amounts of heat transfer and the possibility for surface reaction, in comparison to a radial crack which would rely more on recirculation and diffusion of gases to the interface. The radial orientation is chosen for the following calculations to better represent those cracks which are very narrow in nature. All of the items in Figure 4 are included in the calculations as well as reactions which occur within the shown control volume. The reactants and products for the calculations consisted of H_2 , O_2 , CO , CO_2 , H_2O , SO_2 , N_2 , NO , Cr_2O_3 , Fe_3C , $FeO(s)$, $FeO(l)$, $Fe_3O_4(s)$, $Fe_2O_3(s)$, Fe , $Fe(l)$, $FeO(g)$, FeO_2 , H_2 , NO_2 , and $NI(l)$. Other possible products, such as sulfur-containing species, were not included because the interior ballistic calculation of the charge did not contain sulfur from the ignitor. The evidence produced by Dr. Cote (Cote and Rickard 1999) warrants further investigation of sulfur.

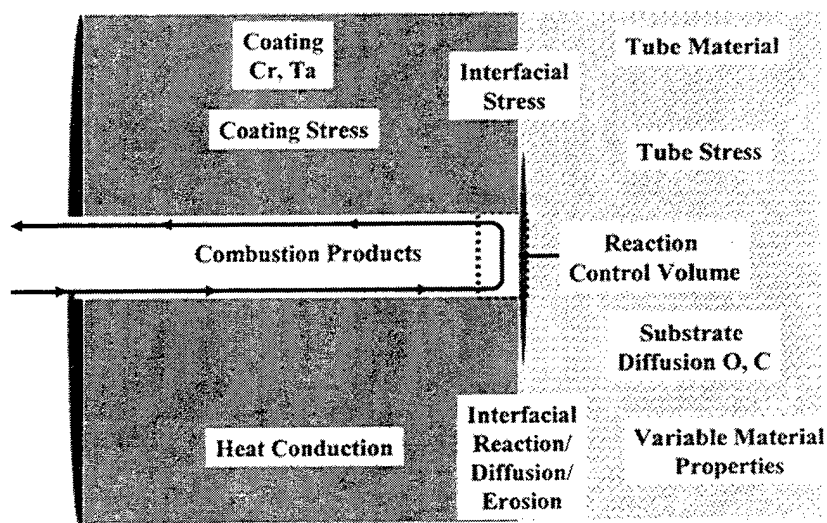


Figure 4. Analytical Surface Coating Description Including Stress.

The results of the erosion calculations are presented in Figure 5 and show a general trend to decrease forward from the forcing cone region. The amount of material removed from the bottom of a crack is approximately $1-1.5 \mu$ per shot. This is an "order of magnitude" given the firing history of 120-mm cannons and subsequent micrographs of subsurface erosion, as shown in Figure 6 and

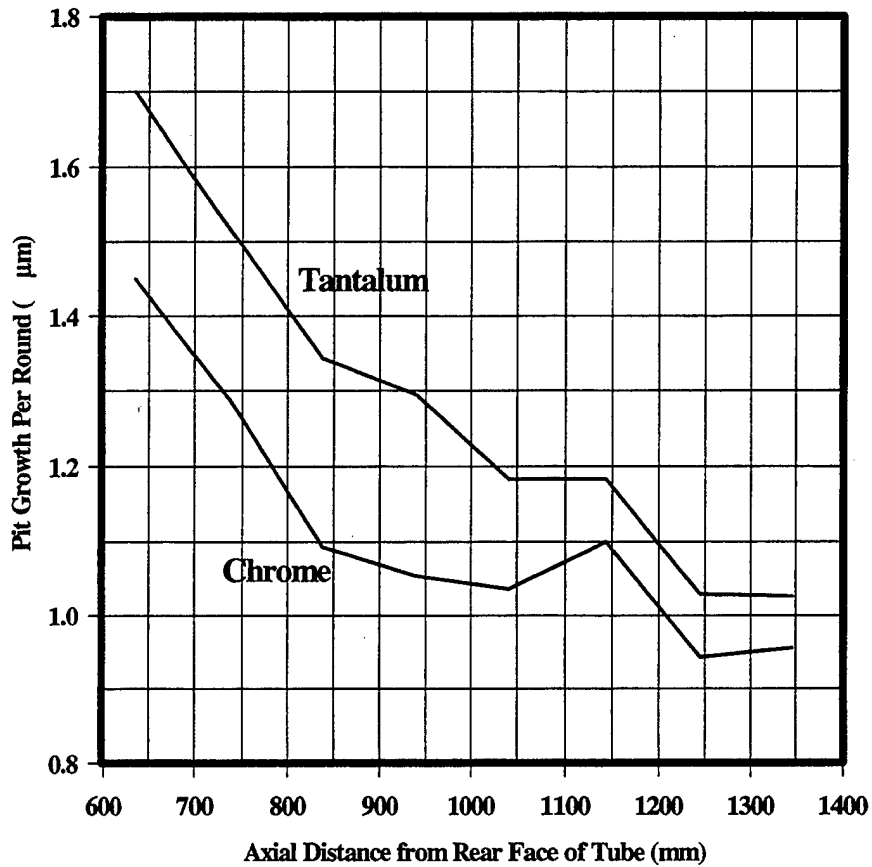
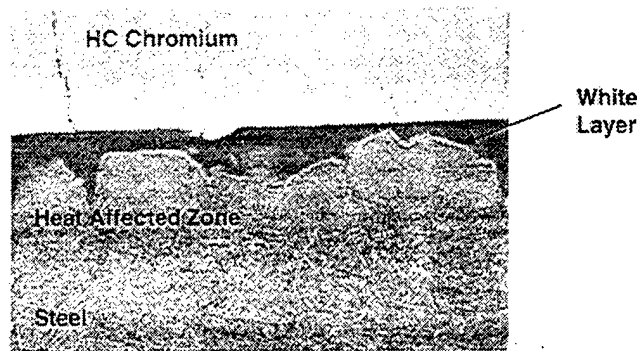


Figure 5. Single-Shot Pit Erosion Depth for 0.010-in Chromium- and Tantalum-Plated Gun Tubes Firing 120-mm APFSDS Rounds.



Original 500X

Figure 6. Substrate Erosion Beneath "Older" Cracks Showing Lateral Distribution (Source: Dr. Paul Cote, Benet Laboratories).

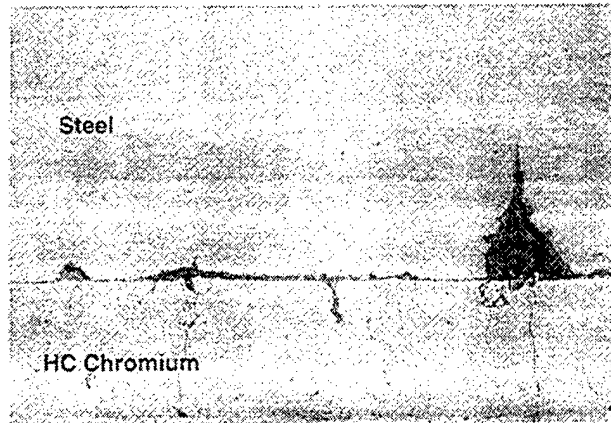
provided by Dr. Cote. Figure 5 also shows that due to thermophysical properties between the coatings, tantalum might have more rapid deterioration at the bottom of a crack than chromium. This confounds the previous tube service life prediction that tantalum tends to crack less than chromium, if at all due to its dilation properties.

The erosion did not predominantly progress downward from the bottom of the crack in the micrograph in Figure 6 but rather traversed along the interfacial boundary, essentially debonding the substrate from the coating. The crack shown may be an older one which may have been formed due to residual stresses from manufacturing. Also important to note from this figure is that the cracks can be a small fracture (narrow) in comparison to the coating thickness. The white layer exhibited in this micrograph indicates high levels of heat input at the substrate surface.

Figure 7 reveals significant pit erosion, possibly produced when a crack first forms. Given more firings on this particular specimen, the sample may ultimately appear as Figure 6 appears. Important to note in Figure 7 is that the interface has been degraded significantly with minimal in-depth erosion in the regions away from the erosion intrusion. This crack, which produced the substrate erosion in Figure 7, is like Figure 3 where it was filled with material—primarily oxides, iron, sulfides, and talc from the charge additive for wear reduction. Iron sulfide melts at 1,463 K, whereas iron carbide melts at 1,423 K. This indicates that iron sulfide is certainly an important factor in the erosion process and has not been included in these calculations.

4. Liner-Substrate Interfacial Degradation Concepts

The degradation of the interface is evident in both Figures 6 and 7. The chemical reactions between the propellant products and the substrate gun steel in this interfacial region are primarily caused by the temperature of the substrate. The subsurface temperature has both convective and conductive contributions, as in equation 32.



Original 500X

Figure 7. Substrate Erosion Beneath “Newer” Cracks Showing Lateral Distribution (Source: Dr. Paul Cote, Benet Laboratories).

$$Q = h(T_{gas} - T_s) - K \frac{\partial^2 T}{\partial x^2} - \frac{K}{r} \frac{\partial^2 T}{\partial r^2}. \quad (32)$$

Without a crack or defect present in a 0.01-in chromium-lined 120-mm M256 tank cannon firing an APFSDS round, the peak interfacial temperature is about 800 K, as shown in Figure 8. If the substrate material is separated from the coating by a gap, then one of the heat transport mechanisms (conduction across the interface) is interrupted. Thus, only indirect and highly inefficient conduction, as well as the convection, can contribute resulting in a much reduced substrate peak temperature below the region of interruption in contact. If, however, the substrate material is in intimate contact with the coating near the degraded interface, then conduction around the edge takes place. Note that if the coating is separated, as in Figure 8, then the coating peak temperature rises dramatically. An important factor to keep in mind is that this differential in peak temperatures between the substrate and coating is influenced by the size of the undermined region. The larger the undermined region, the higher the temperature difference is between the coating material and substrate. Thus, as the service life increases, the rate at which the undermining occurs will probably increase as well.

At least three hypotheses can describe how the chromium is subsequently removed once the coating is sufficiently undermined. The first is the obvious one in which the coating lifts or falls out

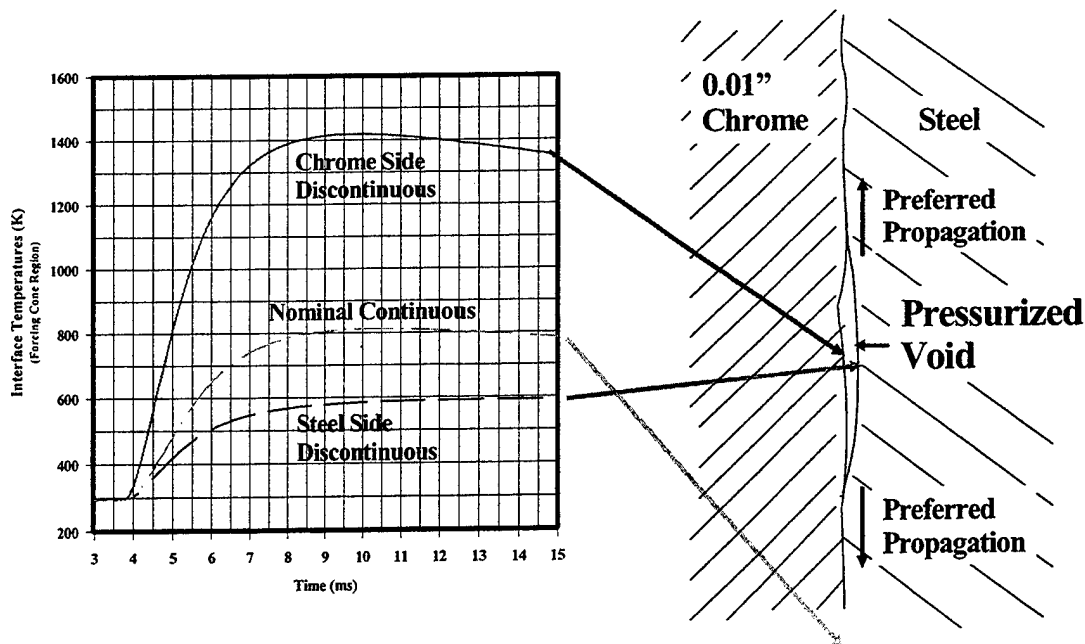


Figure 8. Interfacial Temperatures for Continuous and Discontinuous Regions With Descriptive Diagram.

due to complete separation from the substrate. The second removal method is more probable, novel, and violent. If the cracks surrounding the coating platelet are small enough to enable gas choking to occur during blowdown, the orifice under the platelet would act as a mini-bore evacuator and retain high pressure. This would expel the remaining platelet off the surface in a potentially dramatic fashion during blowdown. Depending upon the ultimate strengths of each, either the substrate or coating would fail in tensile fracture or the interface would fail. The ultimate strength of chromium is about 413 MPa (Matweb 1999), while that of gun steel is about 1,500–1,800 MPa (Holt et al. 1975). This would suggest that bits of chromium might be found on the surface after the platelet is removed if it was not completely undermined. The third method is that during the corrosion process, the developing and expanding oxide layer slowly pushes the chromium upward, enabling projectile engagement.

5. Summary

The principal finding and discussion from this research are that neither chromium nor tantalum coatings (0.010 in thick) fail in axial or radial shear. The coatings crack or plastically deform to relieve stress. The computed platelet size is consistent with experimental observations. Tantalum coatings appear to crack less than chromium coatings. (Hopefully, they will not crack at all.) The computed substrate erosion is consistent with experimental observations. Reaction sets (kinetics) will assist in more accurately describing the surface conditions. Sulfur may be a significant player in the surface reaction process by producing FeS, which melts at 1,463 K. The coating substrate interface is the most likely region of erosion and subsequent failure due to energy transport mechanisms. A literal undermining or unzipping of the interface occurs through thermodynamically driven thermochemistry. Once the coating is undermined, it possibly acts as a pressure vessel after the passage of the rarefaction wave during the blowdown process. This may lead to forced removal of the platelets. Subsequent corrosion may push the platelets from the surrounding coating and enable the projectile passage to remove it. In this line of reasoning, storage conditions may play some role in postfiring oxidation of subsurface exposed steel, and possibly, a postfiring treatment of the gun can be found to control corrosion, which may increase tube life dramatically in the future.

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